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High temperature fatigue properties of a Si-Mo ductile cast iron

P. Matteis[a], G. Scavino[a], A. Castello[b], D. Firrao[a,]*

[a]Politecnico di Torino, DISAT dep., Corso Duca degli Abruzzi 24, It-10129, Torino, Italy
[b]Centro Ricerche FIAT, Group Material Lab, Corso Settembrini 40, It-10135, Torino, Italy

Abstract

Si-Mo ductile cast irons are used to fabricate the exhaust manifolds of the internal combustion engines of large series cars. The maximum pointwise temperature at full engine load may be higher than 750 °C, and the main damage mechanisms in service are high-temperature oxidation and thermo-mechanical fatigue. The mechanical behavior of a Si-Mo cast iron is studied as a function of temperature, by means of stress-life fatigue tests up to 10 million cycles, as well as tensile and compressive tests, followed by fractographic examinations, and the mechanical test results are correlated with both the cast iron microstructure and previous heat treatment.

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Keywords: Si-Mo cast iron; mechanical properties; fatigue

1. Introduction

Si-Mo ductile cast irons, containing 4 - 6 % Si and 0.5 - 2 % Mo, are employed for high temperature applications. The high Si content both increases the high temperature corrosion resistance, by forming a Si-rich surface oxide layer, and stabilizes the ferritic matrix, primarily by increasing the A1 transformation temperature, whereas the Mo addition increases the high temperature strength and improves the creep behavior, by forming alloy carbides in the ferritic matrix (Jenkins et al. 1990, Li et al. 2004, Tholence and Norell 2008, Yoon-Jun Kim et al. 2009).

These alloys are often used to fabricate the exhaust manifolds of the internal combustion engines of large series

* Corresponding author. Tel.: +39-0110904663; fax: +39-0110904699.
E-mail address: donato.firrao@polito.it

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cars (Zeytin et al., 2009). In this application the maximum pointwise temperature at full engine load may be higher than 750 °C, and the main damage mechanisms in service are high-temperature oxidation and thermo-mechanical fatigue, the latter being caused by the engine start and stop and by the variation of its power output. Exhaust manifolds may undergo small-scale yielding in service, e.g. up to 0.4% equivalent plastic strain at full load, thus the thermo-mechanical fatigue occurs in the elastic-plastic regime.

This work is devoted to study the mechanical behavior of the Si-Mo cast irons as a function of temperature, by means of tensile and compressive tests and of high-frequency stress-life fatigue tests, and to correlate it with both the cast iron microstructure and previous heat treatment.

2. Experimental

The cast iron was modified (spheroidized) with Mg-alloy wire, inoculated with Fe-Si powder, and the samples were separately cast in sand molds in the shape of bars with cross-sectional dimensions in the 25 to 50 mm range. The bars average chemical composition was (mass %): C 3.3, Si 4.0, Mo 1.4, Ni 0.49, Cr 0.15, Mg 0.03, P 0.05 and S 0.003. Thereafter, the bars were ferritized, i.e. annealed at 800 °C for 3.5 h and cooled at 0.6 °C/min from 800 to 650 °C and 3 °C/min from 650 to 300 °C; mechanical test specimens were machined from the annealed bars.

The tensile test, with 8 mm diameter and 56 mm calibrated length specimens, as well as the compression tests, up to 1.5 % strain, were performed at different temperatures, with actuator displacement control.

The stress-life tension-compression fatigue tests, with load ratio R = -1, were performed either at room temperature (22 °C) with about 80 Hz frequency, or at 400 and 700 °C with about 150 Hz frequency, by using hourglass specimens, with minimum diameter 9 mm and fillet radius 75 mm (stress concentration factor k_{tn} = 1.05). The maximum nominal stress is here reported as the fatigue stress or strength. The fatigue strength at 10 million cycles was eventually calculated with the staircase method.

In the compression tests, the strain rate was comprised between -5.2E-5 and -3.1E-5 s^{-1}. In the tensile tests the strain rate was about 1.3E-4 s^{-1} at the test start, it was suddenly increased to about 5.9E-4 s^{-1} by increasing the displacement rate after the yield point, and then slightly decreased due to the specimen elongation. The sudden strain rate increase performed after the yield point was used to evaluate the strain rate sensitivity (between 0.05 and 0.5 % true plastic strain). Finally, in the fatigue tests the strain rate was estimated between ±0.5 and ±0.8 s^{-1} in the room temperature tests, and between ±1.5 and ±2.3 s^{-1} in the high temperature ones.

3. Results

The as-cast microstructure consists of graphite, eutectic alloy carbides (i.e., Mo-rich carbides formed during solidification), ferrite, and lamellar pearlite, with rare porosities. The mean size of the graphite spheroids is about 17 μm, the graphite volume fraction is about 10 %, and the spheroidal graphite fraction is about 70 %. The graphite particles are surrounded by the ferrite, whereas the eutectic carbides are distributed on the solidification cell walls, far away from the graphite, and are surrounded by the lamellar pearlite.

After the above described ferritizing anneal (i.e., immediately before all the mechanical tests) the graphite and the eutectic (primary) carbides are sensibly unaffected (Fig 1a), but the lamellar pearlite is transformed into either ferrite or globular pearlite, and the ferrite exhibit small, secondary carbides, about 0.25 μm large (Fig. 1b, c), in both intergranular and intragranular positions. Moreover, the secondary carbides are enriched in Mo.

Finally, a small but detectable microstructural evolution occurs during the tests performed at high temperature for long durations; e.g., after about 20 h at 700 °C the graphite and the eutectic carbides are still unaffected, but the secondary carbides in the ferritic matrix increase in number and size and are further enriched in Mo, and the former pearlitic areas become barely discernible from the former ferritic ones, exhibiting only a slight difference in the secondary carbides size and shape (Fig. 1d).

The tensile and compressive properties of the examined material (in the annealed condition) are shown in Fig. 2 and 3, as a function of temperature. The difference between the yield stress and the ultimate tensile strength, as well as the uniform elongation, generally decrease by increasing the test temperature; in particular, at and above 500 °C the difference between the yield stress and the ultimate tensile strength is very small and may be due mostly to the above mentioned strain rate step (which occurs in between the yield point and ultimate tensile strength).

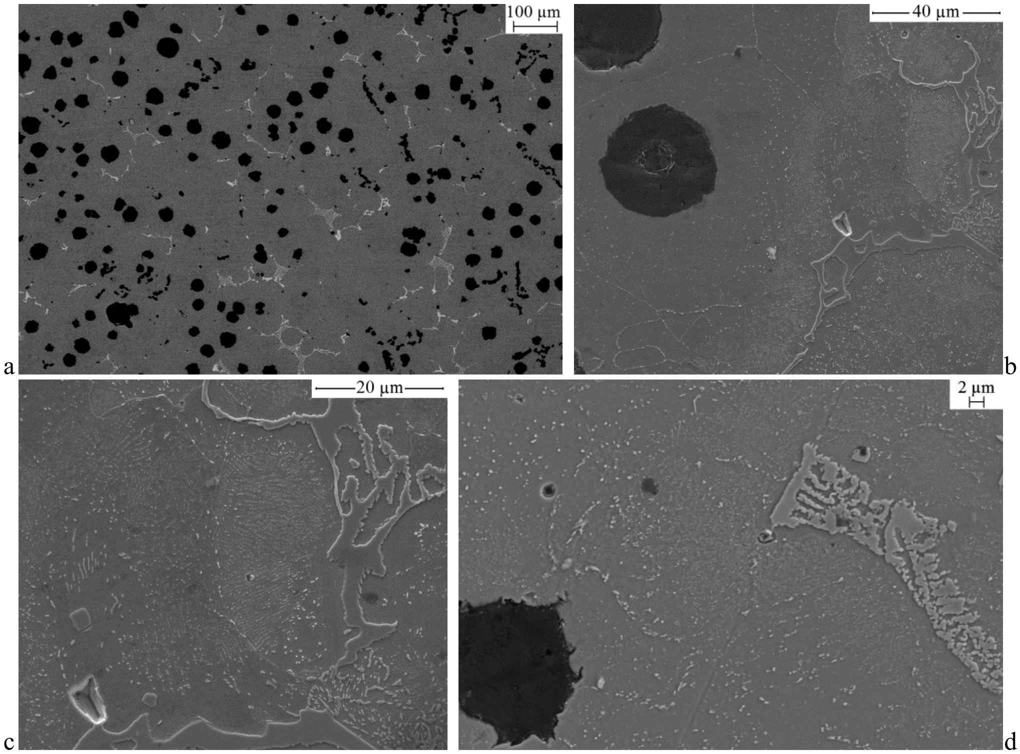

Fig. 1. Microstructure of the Si-Mo cast iron after the 800 °C heat treatment (a, b, c) and after testing at 700 °C for about 20 h (d). Nital etch. Scanning electron microscopy with backscattered electrons (a) and secondary electrons (b, c, d).

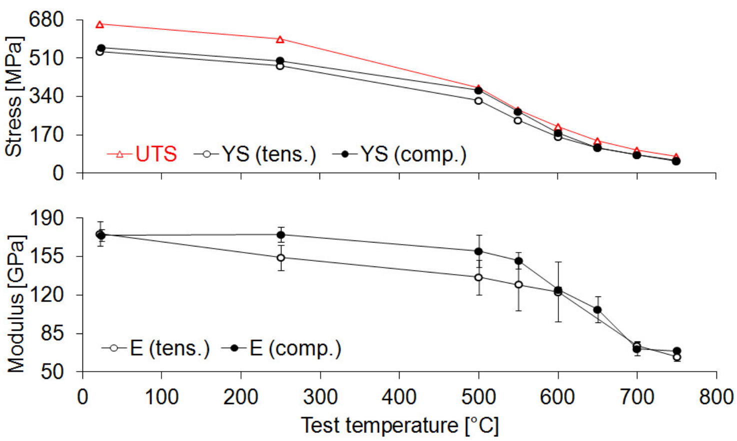

Fig. 2. Tensile and compressive properties of the Si-Mo cast iron as a function of the test temperature: elastic modulus (E), 0.2% yield stress (YS) and ultimate tensile strength (UTS).

The strain rate sensitivity exponent at room temperature is positive but very small (0.003), it is negative at 250 °C (-0.009), and thereafter it becomes positive again and remarkably increases with temperature, up to 0.1 at 750 °C.

The tensile specimens exhibit a slight necking only at temperatures equal to or higher than 550 °C, and their fracture surfaces are overall normal to the tensile axis at all examined temperatures, even if their macroscopic roughness increases by increasing the temperature.

At room temperature the tensile fracture surface is formed by cleavage of the ferritic matrix and detachment of the graphite particles (Fig 4a), and the crack path is almost planar.

By increasing the temperature, the matrix fracture surface exhibit an increasing amount of ductile fracture, and eventually becomes fully ductile at and above 550 °C. The ductile fracture occurs by microvoid coalescence, with large voids originating from the graphite particles and small ones originating from the matrix. By further increasing the temperature, the growth of the large voids is enhanced and the crack path includes an increasing number of decohered graphite particles (Fig 4b).

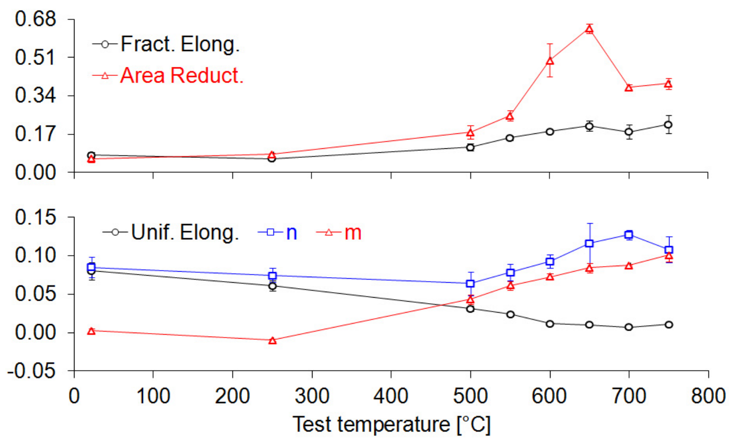

Fig. 3. Tensile properties of the Si-Mo cast iron as a function of the test temperature: fracture elongation, area reduction, uniform elongation, strain hardening exponent (n) and strain rate sensitivity exponent (m); n and m were evaluated between 0.05 and 0.5 % true plastic strain.

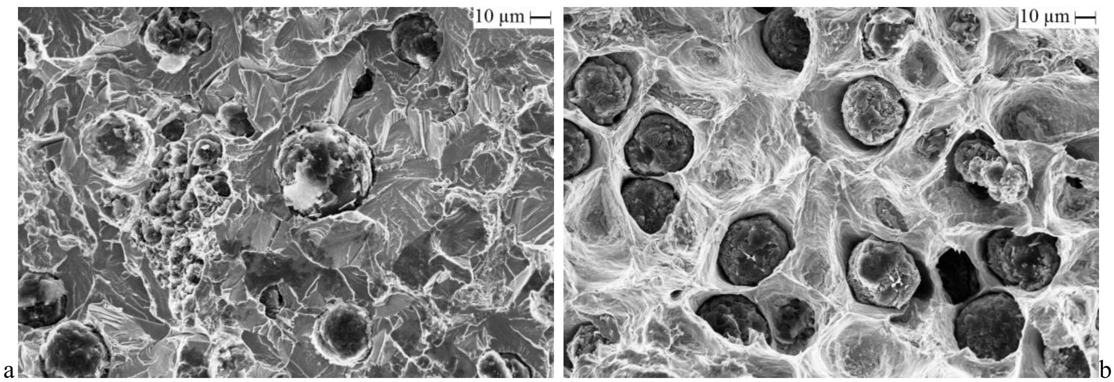

Fig. 4. Fracture surfaces of tensile specimens tested at room temperature (a) and 750 °C (b).

The fatigue strength at 10 million cycles is 256 ± 12 MPa (mean ± standard deviation) at room temperature; it exhibits a barely detectable reduction by increasing the temperature to 400 °C, namely to 247 ± 8 MPa, but it drops to 140 ± 8 MPa by further increasing the temperature to 700 °C (Fig. 5).

The fatigue crack is always nucleated from a shrinkage porosity, due to the casting process, which is either adjacent or close to the specimen outer surface, and the fracture surface is always normal to the tensile axis and close to the minimum section of the hourglass specimens. The fatigue crack growth occurs mainly through the ferritic matrix, where microscopic fatigue striations are formed (Fig. 6), whereas the graphite particles, which are on the fatigue crack path, are detached from the matrix. The eventual overload failure occurs with the same mechanisms that have been observed in the tensile tests, as a function of the test temperature.

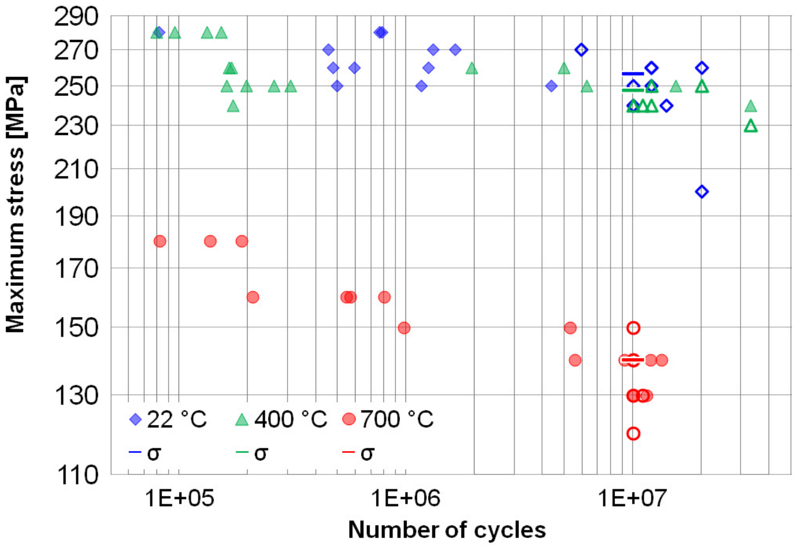

Fig. 5. Fatigue stress-life tests at temperature 20, 400 or 700 °C, with load ratio R = -1, and 10 million cycles fatigue strength (σ) at each temperature level. Not broken specimens are represented by empty symbols.

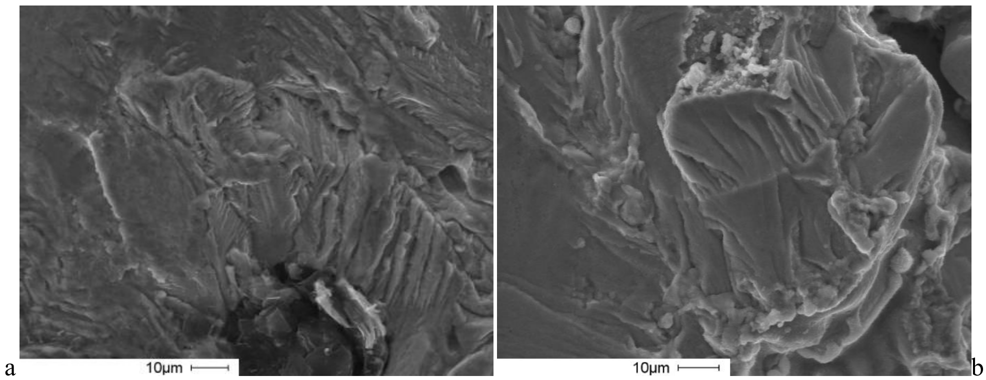

Fig. 6. Fatigue crack growth surfaces of the specimens tested as follows, with load ratio -1: (a) room temperature, maximum stress 260 MPa, broken at 588,000 cycles; (b) 700 °C, 140 MPa, 9,190,000 cycles.

4. Discussion and conclusions

The overall as-cast microstructure of the examined 4 % Si, 1.4 % Mo ductile cast iron consists of spheroidal graphite, eutectic (primary) Mo-rich alloy carbides, ferrite and pearlite, and is compatible with previous observation on Si-Mo ductile iron castings (Yoon-Jun Kim et al. 2009, Zeytin et al. 2009). The subcritical anneal converts the lamellar pearlite into globular pearlite and causes the precipitation of intergranular and intragranular secondary carbides.

The high-temperature tensile test results are consistent with previously known ultimate tensile strength values (Jenkins et al. 1990), and show that, by increasing the temperature, especially above 500 °C, the examined material exhibits a marked decrease of the uniform elongation (down to 1% above 600 °C, from 8% at room temperature) and a large increase in the strain rate sensitivity (up to 0.10, from close to nil at room temperature).

The (high-frequency) 10 million cycles fatigue strength is remarkably constant between room temperature and 400 °C, and thereafter decreases, but not nearly as much as the yield stress and the ultimate tensile stress. In particular, the ratio of the (high-frequency) fatigue strength to the ultimate tensile strength, which is 0.39 at room temperature, increases to about 0.53 at 400 °C, and eventually to 1.38 at 700 °C. The latter value can be explained by considering the very large difference in the strain rate, which is about 4 orders of magnitude larger in the fatigue tests than in the tensile ones, and it is hypothesized that this peculiar behavior at high temperature is correlated with the above noted large increase in the strain rate sensitivity, in the same temperature interval.

The latter hypothesis can be developed as follows. The effective yield stress and ultimate tensile stress values at a given strain rate can be calculated by considering the measured (quasi-static) values, the measured strain rate sensitivity exponent, and the ratio between the given strain rate and the strain rates which were employed in the tensile tests. By using the latter method, in the 700 °C case, the effective yield stress and ultimate tensile stress, corresponding to the strain rate employed in the fatigue tests, are about 210 and 230 MPa, respectively, i.e. they are reasonably larger than the measured fatigue strength (144 MPa), and the ratios between the fatigue strength and the effective yield stress and ultimate tensile stress also become reasonable, namely 0.67 and 0.61.

Therefore, it is concluded that the present results are overall consistent with the hypothesis that the high-temperature fatigue tests are greatly influenced by the strain rate and, ultimately, by the frequency of the testing machine. It must be nevertheless be noted that the latter calculations are extrapolations affected by a large uncertainty, and should hopefully be confirmed by future targeted experiments at different strain rate

Finally, the fractographic examinations allowed to recognize fatigue striations at all temperatures, and showed that the overload failure occurs by cleavage at room temperature, and by ductile fracture at high temperature.

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